Study of Exchange Anisotropy for Ni$_{80}$Fe$_{20}$/Fe$_{60}$Mn$_{40}$ (111) Epitaxial Films

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Deposited 07/18/2019

Citation of published version:

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Citation: Journal of Vacuum Science & Technology A 19, 1213 (2001); doi: 10.1116/1.1340660
View online: https://doi.org/10.1116/1.1340660
View Table of Contents: https://avs.scitation.org/toc/jva/19/4
Published by the American Vacuum Society
I. INTRODUCTION

Currently, the exchange anisotropy is of particular interest due to its "pinning" effect used in giant magnetoresistance spin valve heads. Experimentalists are trying to find an ideal antiferromagnet with outstanding properties, i.e., large enough exchange bias field, small coercivity, good thermal stability, and good corrosion resistance. On the other hand, theorists are investigating the microscopic mechanism of the phenomenon. Many systems have been studied and several models were proposed. Since the exchange anisotropy is an interfacial effect, the antiferromagnet spin configuration at the interface is believed to play a key role. The ferromagnet Ni$_{80}$Fe$_{20}$ is often treated as a single domain in contact with the antiferromagnet. Thus, a strong exchange interaction between the ferromagnetic spins and the antiferromagnetic spins will make the antiferromagnetic spins deviate from their original directions. The situation is further complicated by the intrinsic properties of the antiferromagnet such as its crystal structure, anisotropy, exchange stiffness, nature of spin compensation in an ideally atomically flat plane, and the extrinsic properties such as crystallinity, grain size, grain orientations, and roughness. This makes the quantitative estimation of exchange anisotropy very difficult. To determine the strength of exchange bias, the traditional hysteresis loop method measures the energy difference between two states of the ferromagnet in opposite directions. It involves not only reversible but also irreversible processes, i.e., switching of antiferromagnetic grains driven by the ferromagnet. The loop shift is then not a good estimation of the interfacial exchange coupling strength. A better way to estimate the ferro–antiferromagnetic interfacial exchange coupling strength is the reversible method, i.e., initial susceptibility at a low field transverse to the pinning direction.

Among the variety of antiferromagnetic materials, FeMn maybe one of the most studied metallic antiferromagnetic compounds. Yet, this system is still not well understood due to its complicated magnetic structure and the difficulty of growing single crystal films. Antiferromagnetic $\gamma$-FeMn is a metastable face-centered-cubic (fcc) phase. Even with a conforming buffer layer such as Ta or fcc Ni$_{80}$Fe$_{20}$, FeMn undergoes a transition from the $\gamma$-phase to the $\alpha$-body-centered-cubic (bcc) phase as film thickness is increased beyond several hundred angstroms. Most of the data on antiferromagnet $\gamma$-FeMn is for polycrystals. An exception is the work of Jungblut et al., who grew epitaxial Ni$_{80}$Fe$_{20}$/Fe$_{60}$Mn$_{40}$ exchange bilayers on Cu single crystal substrates by molecular-beam epitaxy and reported their magnetic properties.

We report the epitaxial growth of Ni$_{80}$Fe$_{20}$/Fe$_{60}$Mn$_{40}$ on Si(110) using a Cu buffer layer. The fcc(111) Cu growth on Si(110) was reported by Hai Jiang et al. Detailed information on the structure evolution during growth is presented. The magnetic properties of the exchange bilayers are discussed and a comparison is made with corresponding polycrystalline films.

II. EXPERIMENTAL METHODS

The ultrahigh vacuum chamber used for the film deposition had a base pressure of $10^{-8}$ Torr. Cu/Ni$_{80}$Fe$_{20}$/Fe$_{60}$Mn$_{40}$ were dc sputtered on H–Si(110) with an Ar pressure of 3 mTorr. To avoid contamination of interfaces, there was no waiting time between the deposition of Ni$_{80}$Fe$_{20}$ and Fe$_{60}$Mn$_{40}$. A Cu capping layer of 5 nm was DC sputtered on Fe$_{60}$Mn$_{40}$ to protect it from oxidation in the air. A magnetic
field of ~100 Oe was applied during film deposition to induce the anisotropy axis. Four sets of samples were prepared.

(a) Cu(t)/Ni$_{80}$Fe$_{20}$(10 nm)/Fe$_{60}$Mn$_{40}$(20 nm) with t=0, 1, 2, 3, 5, 10, 20, 30, and 100 nm.
(b) Cu(20 nm)/Ni$_{80}$Fe$_{20}$(10 nm)/Fe$_{60}$Mn$_{40}$(t) with t=10, 20, 30, 40, and 50 nm.
(c) Cu(20 nm)/Ni$_{80}$Fe$_{20}$(5 nm)/Fe$_{60}$Mn$_{40}$(t) with t=2.5, 5, 7.5, and 10 nm.
(d) Cu(20 nm)/Ni$_{80}$Fe$_{20}$(t)/Fe$_{60}$Mn$_{40}$(10 nm) with t=2.5, 5, 7.5, and 10 nm.

To monitor the film crystallinity and structural evolution, some reference samples were prepared with the same thickness and conditions as above. In situ low-energy electron diffraction (LEED) and reflection high energy electron diffraction (RHEED) were performed on the substrate and each layer of the reference samples. A transmission electron microscope (TEM) plane view was performed to observe the grain sizes and film structure of some samples in set (a). X-ray diffraction (XRD) θ-2θ scans and rocking curves were used to detect the structure and perfection of the crystals. Surface roughness was measured by an atomic force microscope (AFM) immediately after the samples were taken out of the vacuum chamber.

Hysteresis loops were measured by magneto-optic Kerr effect (MOKE) for films with thin Fe$_{60}$Mn$_{40}$ layers (<20 nm) and by vibrating sample magnetometer (VSM) for films with thicker Fe$_{60}$Mn$_{40}$ because the MOKE signal was too weak to be detected through these thicker layers. The agreement of these two measurements was confirmed by comparing results for films with thin Fe$_{60}$Mn$_{40}$. To avoid the training effect (see Ref. 9 in Ref. 13, Schlenker et al.), before measurement of every sample 20 loop cycles were done along the easy axis with a maximum field larger than the saturation field of the ferromagnetic layer. To obtain more information of the anisotropy, hysteresis loops were measured at different in-plane angles with a step of 15°.

To perform the reversible measurement, the anisotropy axis was first determined by rotating the sample in-plane and finding the angle of maximum coercivity. A field of several hundred oersteds was applied in the pinned direction to minimize the dispersion of anisotropy axes. The sample was then rotated 90° and a small ac field of amplitude 5 Oe and period of several seconds was applied to measure the initial susceptibility along the hard axis. Extrapolating the obtained straight line to the saturation moment measured from the easy axis thus gave the magnitude of the exchange anisotropy.

III. EXPERIMENTAL RESULTS

A. Structural analysis

1. Films with varying Cu buffer layer thickness but fixed thickness of other layers, i.e., film set (a)

Fcc(111) Cu growth on Si(110) was confirmed by the LEED patterns which showed six fold symmetry, except for the 1 nm Cu film where no LEED pattern was observed. For Cu thinner than 20 nm, diffuse spots were observed, a result of stress due to lattice mismatch. The diffraction spots became sharper with increasing Cu thickness suggesting the stress is reduced and the crystallinity improved.

Ni$_{80}$Fe$_{20}$ films directly deposited on Si(110) did not show LEED patterns, and RHEED verified the film was polycrystalline with only diffraction rings observed. Very faint LEED pattern was observed for Ni$_{80}$Fe$_{20}$ on 1 nm Cu buffer layer, suggesting the stress still remained although it was released to some extent. RHEED indicated a mixture of single crystals and polycrystals for this film and the ring intensity was stronger than the spots. For the films with Cu buffer layer thickness larger than 1 nm, clear LEED patterns were observed on the Ni$_{80}$Fe$_{20}$ layer. As expected, films deposited on thick Cu buffer layer showed better growth. When the Cu buffer layer thickness was increased to 3 nm, the portion for polycrystals in Ni$_{80}$Fe$_{20}$ decreased greatly and the diffraction rings became very weak. The Ni$_{80}$Fe$_{20}$ was single crystal when deposited on thicker Cu buffer layers and no diffraction rings were observed. These results show the importance of the crystal quality of the underlying Cu layer. Namely, the crystallinity of Ni$_{80}$Fe$_{20}$ followed that of Cu. Figure 1 shows RHEED patterns for Si(110)/Cu(t)/Ni$_{80}$Fe$_{20}$(10 nm) with Cu thickness t=0, 1, 3, 10, and 100 nm. The improvement of crystallinity is obvious.

No LEED patterns were observed for Fe$_{60}$Mn$_{40}$ films deposited on Ni$_{80}$Fe$_{20}$/Cu. However, RHEED diffraction spots were observed, confirming the fcc(111) epitaxial growth. No difference was found in the RHEED patterns of Fe$_{60}$Mn$_{40}$ with different Cu buffer thickness as long as the Fe$_{60}$Mn$_{40}$ layer thickness was greater than 10 nm, indicating stress was totally released in Fe$_{60}$Mn$_{40}$ layer at this thickness. More importantly, unlike Ni$_{80}$Fe$_{20}$ no rings were found in RHEED patterns of Fe$_{60}$Mn$_{40}$ with thin Cu buffer layers, suggesting that the Fe$_{60}$Mn$_{40}$ films were single crystals. A Cu buffer layer as thin as 1 nm was thick enough to induce a good
epitaxy of Fe\textsubscript{60}Mn\textsubscript{40}. The crystal quality of Ni\textsubscript{80}Fe\textsubscript{20} did not affect the Fe\textsubscript{60}Mn\textsubscript{40} epitaxial growth.

The epitaxy and grain sizes of these films were further studied by TEM. Figure 2 shows the in-plane electron diffraction patterns with a beam direction of Cu\textsubscript{111} and corresponding bright field images for Si(110)/Cu\textsubscript{t}/Ni\textsubscript{80}Fe\textsubscript{20}(10 nm)/Fe\textsubscript{60}Mn\textsubscript{40}(20 nm), with \( t = 1 \) and 100 nm. Consistent with the results from RHEED, ring patterns with spots drastically elongated along the circumference were observed for the \( t = 1 \) nm Cu sample while a spot pattern was found for the \( t = 100 \) nm sample. In Fig. 2(a), the six symmetrical bright arcs in the outer ring result from \{220\} diffraction from either fcc Cu, Ni\textsubscript{80}Fe\textsubscript{20}, or/and Fe\textsubscript{60}Mn\textsubscript{40}, revealing a strong in-plane texture of grains with fcc [111] orientation. There was a substantial improvement in epitaxy when the Cu buffer layer thickness was increased to 100 nm. The \{220\} diffraction arcs in Fig. 2(a) now become relatively sharp spots in Fig. 2(b), confirming that the distribution of in-plane orientations is much more restricted. Individual grains could be identified in the bright field image of the film with a 1 nm Cu buffer layer, as shown in Fig. 2(c). Dark field images of this sample (not shown here) reveal that the grain size ranges from 1 to 10 nm with an average of approximately 5 nm, much smaller than the film thickness. The contrast seen in the film with the 100 nm Cu buffer layer is not associated with individual grains but due to defects (dislocations, strain, etc.).

From the XRD high-angle \( \theta-2\theta \) scans of the films, we found that the Cu\textsubscript{111} + Fe\textsubscript{60}Mn\textsubscript{40}\textsubscript{111} and Ni\textsubscript{80}Fe\textsubscript{20}\textsubscript{111} peaks dominated the spectrum. Since polycrystalline Cu and Ni\textsubscript{80}Fe\textsubscript{20} were found in films with very thin Cu buffer layers, it was concluded these polycrystals were textured with the [111] direction normal to the film plane. The peak intensity for Cu\textsubscript{111} + Fe\textsubscript{60}Mn\textsubscript{40}\textsubscript{111} increased and the width for the Cu\textsubscript{111} + Fe\textsubscript{60}Mn\textsubscript{40}\textsubscript{111} peak decreased with the increase of Cu buffer thickness. We also measured the rocking curves for Cu(111) + Fe\textsubscript{60}Mn\textsubscript{40}\textsubscript{111} of Si(110)/Cu\textsubscript{t}/Ni\textsubscript{80}Fe\textsubscript{20}(10 nm)/Fe\textsubscript{60}Mn\textsubscript{40}(20 nm)/Cu(5 nm) with \( t = 1, 10, \) and 100 nm.

The full width at half maximum decreased from 3.3° for \( t = 1 \) nm to 1.1° for \( t = 100 \) nm and the rocking curve became much sharper, indicating strain was relieved for thick Cu buffer layers.

### 2. Films with varying Fe\textsubscript{60}Mn\textsubscript{40} thickness but fixed thickness of other layers, i.e., film set (b)

The evolution of FeMn phase with thickness was investigated by Kung et al.\textsuperscript{17} They found the FeMn \( \gamma \) phase was metastable and changed to the stable nonantiferromagnetic \( \alpha \)-bcc phase when the film thickness was larger than 36 nm. The films they studied were polycrystalline. It would be interesting to see the phase transition in single crystals. In our experiment, RHEED was employed to study the phase transition of epitaxial Fe\textsubscript{60}Mn\textsubscript{40} film. One great advantage of RHEED is its \textit{in situ} detection of film structure during growth, enabling the observation of the stability of the \( \gamma \) phase with thickness. Diffraction from single crystals creates spots with certain symmetry, depending on the structure. For polycrystalline films, diffraction rings appear due to the random orientation of individual grains. If the polycrystal has the same structure as the single crystal, rings will coincide with the spots. Thus single crystals can be judged from the diffraction spots while the phase transition can be judged from the change of the radii of the diffraction rings. Figure 3 shows the RHEED patterns of Si(110)/Cu(20 nm)/Ni\textsubscript{80}Fe\textsubscript{20}(10 nm)/Fe\textsubscript{60}Mn\textsubscript{40}\textsubscript{t} with \( t = 20, 40, 60, \) and 100 nm. We found no difference between the RHEED patterns for 10 and 20 nm Fe\textsubscript{60}Mn\textsubscript{40} and only diffraction spots were found, indicating the \( \gamma \) phase was stable below 20 nm thick, and the growth was epitaxial. However, as Fe\textsubscript{60}Mn\textsubscript{40} thickness is increased the diffraction spots become faint and diffraction rings are superimposed on the spots. The appearance of diffraction rings and their coincidence with spots suggested that the film became polycrystalline but still has fcc structure. As the Fe\textsubscript{60}Mn\textsubscript{40} thickness is increased, the spots became weaker and rings became stronger. With the Fe\textsubscript{60}Mn\textsubscript{40} thickness increased to 60 nm, the RHEED pattern was dominated by rings, indicating that at this thickness, the majority of the film was polycrystalline. However, the radii of the rings did not change significantly with further increase of Fe\textsubscript{60}Mn\textsubscript{40} thickness.
not change, indicating the film was still fcc. As the thickness
was further increased to 100 nm, the radii of the diffraction
rings change indicating that a phase transition occurred.
These data show that the phase transition started at 60 nm
and was completed around 100 nm. The transition from
single crystal \( \gamma\)-Fe\(_{60}\)Mn\(_{40}\) to the \( \alpha\) phase was mediated by a
polycrystalline \( \gamma\)-Fe\(_{60}\)Mn\(_{40}\) layer of about 30 nm thickness.
This thickness is consistent with the 36 nm in Ref. 17.

**B. Surface morphology**

AFM images of the film surface showed that the root-
mean-square (rms) roughness and correlation length in-
creased steadily with Cu buffer layer thickness. As the Cu
buffer layer thickness was increased from 1 to 100 nm, the
rms roughness increased from 0.56 to 1.1 nm and the in-
plane correlation length increased from 23.5 to 47 nm.\(^1\)

**C. Magnetic measurements**

For sample sets (c) and (d), the exchange bias field and
coercivity were consistent with Ref. 13. Namely, the pining
effect appeared when the Fe\(_{60}\)Mn\(_{40}\) thickness was larger than
2.5 nm and the exchange bias field saturated at around 7.5
nm. The exchange bias field and coercivity were inversely
proportional to the thickness of Ni\(_{80}\)Fe\(_{20}\) as long as the
Ni\(_{80}\)Fe\(_{20}\) thickness was larger than 2.5 nm.\(^1\) These results
confirmed the interfacial nature of exchange anisotropy.

Most authors only reported the exchange bias fields for
the FeMn thickness less than 20 nm,\(^1\) because of the insta-
ribility of the FeMn \( \gamma\) phase above this thickness. However,
results for thick antiferromagnets are important in that they
may give some useful information about the antiferromag-
netic domain configurations and can be used to test different
models.\(^3\)\(^-\)\(^9\) From these results, the Fe\(_{60}\)Mn\(_{40}\) was still in the \( \gamma\)
phase below 60 nm thick. With this in mind, we measured
the exchange bias field \( H_{eb} \) of sample set (b), i.e., Cu(20
nm)/Ni\(_{80}\)Fe\(_{20}\)(10 nm)/Fe\(_{60}\)Mn\(_{40}\)(t) with \( t=10, 20, 30, 40, \) and
50 nm. \( H_{eb} \) ranged from 63 to 82 Oe without a systematic
change. Malozemoff’s predictions of monotonically decreas-
ing exchange bias field with antiferromagnet thickness and
the two critical thicknesses\(^1\)\(_8\) were not found.

Figure 4 shows the exchange bias fields \( H_{eb} \) for Cu(1)/
Ni\(_{80}\)Fe\(_{20}\)(10 nm)/Fe\(_{60}\)Mn\(_{40}\)(20 nm) with \( t=1, 2, 3, 5, 10, 20, \) 30, and 100 nm. We also made a sample without Cu buffer
layer, where we found the film was polycrystalline and \( H_{eb} \)
was only 50 Oe, lower than \( H_{eb} \) of single crystal films. For
the single crystal films, from Fig. 4 we found \( H_{eb} \) was around
70 Oe when the Cu buffer layer thickness was more than 5
nm, less than the \( H_{eb} \) of around 90 Oe for the films with
thinner Cu buffer layer. There are two possible explanations
for this behavior. One possibility is the increase of roughness
with Cu buffer thickness. Another possibility is the change
of intrinsic magnetic properties of Fe\(_{60}\)Mn\(_{40}\) due to the struc-
ture improvement. In Table I, a collection of structure data
and magnetic properties is listed for the films with varying
Cu buffer layer thickness.

The exchange anisotropy for films with different Cu
buffer layer thickness was measured both by the reversible
method and the irreversible method. Comparison of the re-
results by using the two methods can give information about
the magnetization reversal process.\(^1\)\(_2\) For comparison, Table
II lists some values from the two measurements. The ex-
change anisotropy, \( H_{p0} \), using the reversible method was
larger than that from the irreversible method, i.e., the shift,
\( H_{eb} \), of hysteresis loops. The difference was not so large
when the Cu buffer layer was thinner than 30 nm. Actually,
\( H_{p0}=H_{eb}+H_c \) \( \rightarrow \) with \( H_c \) the enhanced coercivity.\(^1\)\(_2\) How-
ever, for the films with thicker Cu buffer layer, a remarkable
difference of 80 Oe between the two measurements was
found and the relationship \( H_{p0}=H_{eb}+H_c \) did not hold.

![Image](image_url)

**FIG. 4.** Exchange bias fields \( H_{eb} \) for Cu(1)/Ni\(_{80}\)Fe\(_{20}\)(10 nm)/Fe\(_{60}\)Mn\(_{40}\)(20 nm) with \( t=1, 2, 3, 5, 10, 20, 30, \) and 100 nm.

<table>
<thead>
<tr>
<th>Cu thickness (nm)</th>
<th>0</th>
<th>1</th>
<th>3</th>
<th>10</th>
<th>30</th>
<th>100</th>
</tr>
</thead>
<tbody>
<tr>
<td>rms roughness (nm)</td>
<td>0.56</td>
<td>0.49</td>
<td>0.67</td>
<td>0.89</td>
<td>1.1</td>
<td></td>
</tr>
<tr>
<td>Correlation length (nm)</td>
<td>23.5</td>
<td>25.4</td>
<td>35.2</td>
<td>31.3</td>
<td>47.0</td>
<td></td>
</tr>
<tr>
<td>LEED on Cu</td>
<td>N</td>
<td>Y</td>
<td>Y</td>
<td>Y</td>
<td>Y</td>
<td>Y</td>
</tr>
<tr>
<td>LEED on Ni(<em>{80})Fe(</em>{20})</td>
<td>N</td>
<td>Y</td>
<td>Y</td>
<td>Y</td>
<td>Y</td>
<td>Y</td>
</tr>
<tr>
<td>LEED on Fe(<em>{60})Mn(</em>{40})</td>
<td>N</td>
<td>N</td>
<td>N</td>
<td>N</td>
<td>N</td>
<td>N</td>
</tr>
<tr>
<td>RHEED on Cu</td>
<td>ST</td>
<td>ST</td>
<td>ST</td>
<td>ST</td>
<td>ST</td>
<td>ST</td>
</tr>
<tr>
<td>RHEED on Ni(<em>{80})Fe(</em>{20})</td>
<td>R</td>
<td>R&lt;SP</td>
<td>SP</td>
<td>SP</td>
<td>SP</td>
<td>SP</td>
</tr>
<tr>
<td>RHEED on Fe(<em>{60})Mn(</em>{40})</td>
<td>R</td>
<td>SP</td>
<td>SP</td>
<td>SP</td>
<td>SP</td>
<td>SP</td>
</tr>
</tbody>
</table>

| Exchange bias \( H_{eb} \) (Oe) | 50  | 94  | 88  | 63  | 69  | 72  |

**Table I.** Collection of rms surface roughness, correlation length, LEED (RHEED) patterns on Si(110)/Cu, Si(110)/CuNi\(_{80}\)Fe\(_{20}\), Si(110)/Cu/Ni\(_{80}\)Fe\(_{20}\)/Fe\(_{60}\)Mn\(_{40}\), and exchange bias fields for Si(110)/Cu(Ni\(_{80}\)Fe\(_{20}\)(10 nm)/Fe\(_{60}\)Mn\(_{40}\)(20 nm) with \( t=0, 1, 3, 10, 30, \) and 100 nm. Note the follow-
ing abbreviations. N: no LEED pattern observed, Y: sharp LEED pattern
observed, Faint: LEED pattern observed but very faint, ST: streaks, R: rings,
SP: spots, R<SP: rings and spots coexist with stronger ring intensity, and
R<SP: rings and spots coexist with weaker ring intensity.

**Table II.** \( H_{eb} \) and \( H_{p0} \) for Cu(1)/Ni\(_{80}\)Fe\(_{20}\)(10 nm)/Fe\(_{60}\)Mn\(_{40}\)(20 nm). \( H_{eb} \) was the loop shift obtained from the irreversible method and \( H_{p0} \) was ob-
tained from the reversible method.
Similar curves were obtained for epitaxial films by other authors.20

Angular dependence showed three maxima and three minima over $360^\circ$ for the film with a 100 nm Cu buffer layer, while two angular curves fit the cosine function. The angular dependence from the Cu surface could be observed only when its thickness was larger than 3 nm. Below this thickness, the RHEED pattern showed streaks superimposed on rings, indicating a mixture of single crystals and polycrystals. Above 3 nm, only diffraction spots were observed, suggesting epitaxial growth.

**B. Coercivity**

We found that for the films with a Cu buffer layer thickness greater than 30 nm the coercivity was much larger than that of films with thinner Cu buffer layers.16 This result suggested that the magnetization reversal process was not dominated by coherent rotation in these samples.22 To find the reasons for the large coercivity and the measurement dependence of exchange anisotropy, we performed magnetic force microscope (MFM) measurements. In all films, we did not find domains or wall motion during magnetization reversal. However, magnetization ripples were observed in the films with a thick Cu buffer layer, while no ripple was found in the films with a thin Cu layer. Magnetization ripple is caused by the dispersion of local anisotropy in films. Ripples in exchange biased films were studied by Egelhoff.23 They found that the strength of ripple patterns was correlated with the coercivity. Our results indicated that significant local anisotropy existed in the films with a thick Cu buffer layer which greatly reduced the initial susceptibility in the transverse direction and greatly increased coercivity. The local anisotropy was not caused by the demagnetizing field due to surface roughness, since the length scale of the ripples was on the order of micrometers (for example, the length scale of the ripples was about 2 $\mu$m for Cu(100 nm)/Ni$_{80}$Fe$_{20}$(10 nm)/Fe$_{60}$Mn$_{40}$(20 nm) and the correlation length of the surface roughness was only tens of nanometers.

**C. An explanation for the different angular behaviors of $H_{eb}$ for epitaxial films and polycrystalline films**

Several authors19,20,24–26 studied angular dependence of exchange bias experimentally and theoretically. Their results showed some unique properties of the exchange coupled systems. Some authors24 attributed the angular behavior to the form of the microscopic interactions between the ferromagnetic and the antiferromagnetic sublattice moments. While others fit the angular curve by including uniaxial or fourfold anisotropy in the ferromagnetic layer.19,20 Based on a magnetizations calculation, Kim et al.25 showed the importance of spin canting of the antiferromagnet on the angular behavior of exchange bias, particularly with the presence of defects. As was shown by Kim et al.,25 a noncosine curve of $H_{eb}$ with angle results from the dispersion of antiferromagnet anisotropy axes. To understand the difference in the angular be-
behavior of $H_{eb}$ for polycrystalline films and epitaxial films, we need to know the origin of the anisotropy of the antiferromagnet for these two kinds of films. In the polycrystalline films, the crystalline anisotropy of individual grains cancels each other due to their random orientations. The anisotropy was mainly due to the stress originating from the lattice mismatch and was uniaxial. On the other hand, for the Fe$_{60}$Mn$_{40}$ epitaxial film the crystalline anisotropy contributes to the total anisotropy. In the fcc(111) Fe$_{60}$Mn$_{40}$ epitaxial layer, the crystalline anisotropy is triaxial because of the six fold symmetry of the crystal structure. The triaxial anisotropy in Fe$_{60}$Mn$_{40}$ is the origin of the dispersion or canting of the spins, yielding the complicated angular behavior of $H_{eb}$. With the increase of the Cu buffer layer thickness, the film crystallinity improved and the stress in the Fe$_{60}$Mn$_{40}$ decreased, resulting in a weaker uniaxial anisotropy, or a relatively stronger triaxial anisotropy. Thus, the spin canting of Fe$_{60}$Mn$_{40}$ increased and the angular behavior of $H_{eb}$ deviated more from a cosine shape. The appearance of three maxima and three minima of $H_{eb}$ versus angle for the film with 100 nm Cu compared to the two maxima and two minima in films with thinner Cu reveals a relatively stronger triaxial anisotropy in the Fe$_{60}$Mn$_{40}$ for the former. The spin canting of Fe$_{60}$Mn$_{40}$ was also the origin of local anisotropy within Ni$_{80}$Fe$_{20}$ since there was a strong exchange coupling between Ni$_{80}$Fe$_{20}$ and Fe$_{60}$Mn$_{40}$. One consequence of the stronger uniaxial anisotropy in Fe$_{60}$Mn$_{40}$ is that fewer Fe$_{60}$Mn$_{40}$ grains switch upon the reversal of Ni$_{80}$Fe$_{20}$, resulting in a smaller $H_{c}$. $H_{eb}$ is enhanced from those Fe$_{60}$Mn$_{40}$ grains with their net moment at the interface fixed upon the reversal of Ni$_{80}$Fe$_{20}$, and $H_{c}$ is enhanced from those that switch. $^4$

V. CONCLUSIONS

The detailed growth and structure of Ni$_{80}$Fe$_{20}$/Fe$_{60}$Mn$_{40}$ on Si(110)/Cu(111) was investigated. The large lattice misfit between Si and Cu was the main source of the stress in the film. Increasing Cu buffer layer thickness released the stress and resulted in a better crystallinity for the exchange bilayer. With the increase of Fe$_{60}$Mn$_{40}$ layer thickness, the metastable fcc $\gamma$-FeMn single crystal films evolved into the stable bcc $\alpha$-phase polycrystalline films, mediated by a fcc polycrystal layer, with the phase transition occurred around the thickness of 60 nm. Film surface roughness increased from 0.56 to 1.1 nm as the Cu buffer layer thickness increased from 1 to 100 nm. The exchange anisotropy measured by using the reversible method and the irreversible method agreed for the films with Cu buffer layer thickness less than 30 nm but showed a remarkable difference of $\sim$70 Oe for the films with larger Cu buffer layer thickness. For the films with Cu buffer layer thickness larger than 30 nm, the coercivity was larger than 30 Oe and magnetization ripple was found. The angular behavior of exchange bias field for epitaxial films deviated from sinusoidal form, while a perfect sinusoidal behavior of exchange bias field was found for polycrystalline films. 

ACKNOWLEDGMENT

This work was supported by NSIC-EHDR Heads Program, ARO Grant No. DAAA-04-96-1-0316 and NSF MRSEC Grant No. DMR-9809423.

$^{23}$W. F. Egelhoff (private communications).